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The brittle fracture of grey cast irons

A. G. GLOVER*† and G. POLLARD*
*University of Leeds, Metallurgy Department, Leeds 2, Yorks.
†Now at CEGB Engineering Laboratories, Marchwood, Hampshire.

Summary

The effect of heat treatment on the fracture of flake cast irons has been studied. The amount of ferrite present in the structure was varied by increasing the time of annealing above or below the pearlite breakdown temperature. Testing was carried out using an instrumented four-point slow bend rig and the fracture surface studied using replicas and a scanning electron microscope.

In all these materials fracture was initiated by cleavage of the graphite flakes ahead of the macroscopic crack front. Subsequent fracture of the matrix depended on the orientation of the pearlite with respect to the crack, the percentage of ferrite and whether or not it was associated with the graphite or pearlite, and the nature of the heat treatment.

The small amounts of ferrite introduced on short heat treatments consisted of a very thin region around the graphite flakes. The K_{Ic} value increased for small additions of ferrite but decreased as the size of the ferrite region became greater than the pearlite interlamellar spacing i.e. at concentrations of approximately 15% ferrite.

Introduction

It has been accepted for many years that the fracture mechanisms of flake cast irons in both the as cast and annealed conditions is controlled by the brittle behaviour of the graphite flakes [1, 2, 3]. The amount, size, shape and distribution of the graphite in grey cast iron greatly influence its properties. The graphite flakes interrupt the continuity of the matrix and also introduce notches at their apexes [4]. Intensive studies have been made on the influence of the chemical composition [5], the rate of cooling of the casting in the mould [6] and the type of graphite formed [7], on the impact fracture characteristics of flake irons.

Recent developments have occurred in the observation of fracture surfaces by the electron microscopy of replicas and the direct observation with scanning electron microscopes [8]. At the same time the analysis and testing of high strength materials has developed with the increased interest in fracture toughness [9, 10].

In this paper results are reported of the study of the fracture behaviour of flake cast irons using microfractographic methods and the application of fracture toughness testing to these low strength, brittle materials.

Experimental procedure

Two basic irons were investigated; a standard flake cast iron, cast into 10 in × 1·2 in dia. test bars, with a mean graphite flake of 0·1 mms., and a large flake cast iron, cast into 10 in × 3·0 in dia. test bars, with a mean graphite flake of 0·2 mms. The analysis of the two irons is given below:

	T.C.%	Si%	Mn%	S%	P%
Standard flake	3·35	2·06	0·46	0·055	0·038
Large flake	3·32	2·08	0·46	0·052	0·038

Heat treatment of a flake cast iron results in the breakdown of the pearlite and deposition of graphite on the existing graphite flakes [11]. This breakdown may be accomplished by heating for prolonged periods below the critical temperature or heating above this temperature. Both types of heat treatment were carried out in a muffle furnace, the details of which are given below:

Material	*Heat treatment*	*% Ferrite*
A. Standard flake:	Held at 800°C for 5 mins, furnace cool after holding at 690°C for 1 hour.	13·2%
B. Standard flake:	Ditto but held at 800°C for 15 mins.	59·6%
C. Standard flake:	Ditto but held at 800°C for 20 mins.	68·1%
D. Standard flake:	Ditto but held at 800°C for 40 mins.	78·0%
E. Standard flake:	Held at 900°C for 4 hours, followed by a furnace cool at 690°C. Held at 690°C for 8 hours before finally furnace cooling (commercial heat treatment).	100%
F. Large flake:	Held at 700°C for 30 mins. Air Cool	1·0%
G. Large flake:	Held at 700°C for 40 mins. Air Cool	4·1%
H. Large flake:	Held at 700°C for 70 mins. Air Cool	26·4%
I. Large flake:	Held at 700°C for 90 mins. Air Cool	40·7%

These treatments resulted in a region of ferrite being produced adjacent to the graphite flakes, the proportion of ferrite being controlled by the heat treatment conditions.

The plane strain fracture toughness parameter K_{Ic} was determined from four point bend specimens. The rig incorporated roller bearings for the outer supports to reduce frictional effects. Crack opening displacements were measured using a double cantilever displacement gauge. Loading was by means of an Instron testing machine with a cross head speed of 0·2 mm/min.

The specimens were between 2·5 and 10 by 10 mm in cross section with a constant span distance of 80 mm. In all cases the results showed that

fracture occurred under plane strain conditions using the ratio 0·5 to 1·0 from the linear region of a plot of apparent K_{Ic} and $B/(K_Q/\sigma_{ys})^2$ against thickness (B).

The value of K_{Ic} was calculated using the method of Srawley [12] modified by Srawley, Jones and Brown [9].

The fracture surfaces of both the slow bend and Charpy impact specimens were examined using a plastic-shadowed carbon replica and by direct observation on a Cambridge Stereoscan microscope. Fracture profiles were studied after nickel plating and metallographic polishing.

Experimental

The room temperature mechanical properties of all the grey cast irons tested are summarised in Table 1. The heat treatment reduces considerably the tensile strength and it can be seen that the plane strain fracture toughness value (K_{Ic}) increases steadily with increase in hardness etc., on heat treating. The variation of K_{Ic} with hardness is shown in Fig. 1 It can be seen that in contrast to the results normally obtained from an impact test (Table 1), the as cast material had a significantly higher value of toughness than the fully heat treated material under the conditions of slow crack propagation. Furthermore when a small proportion of ferrite was introduced into the structure the K_{Ic} value reached a maximum before decreasing to the value of the fully annealed material.

Although fatigue crack initiation was carried out on a number of specimens subsequent testing showed that the fracture toughness value obtained was unaffected and further tests on these materials were performed without the initial fatigue damage.

Microfractography

No distinction could be made between the specimens broken under slow bend and impact conditions with regard to their fracture surfaces. All the materials contained flake graphite and this constituted a major portion of the fracture surface in all the irons. In general, cleavage of the graphite flake took place ahead of the macroscopic crack front as observed in specimen sections removed from the rig prior to failure as shown in Fig. 2. Cleavage of the graphite flakes occurs along their basal planes (0001), into a series of distorted hexagons and short sections of prismatic fracture link the islands of cleavage within the flakes (Fig. 3).

Fracture of the pearlite matrix was observed in two forms. When the crack cut across the lamellae, a typical etched structure of the pearlite was obtained. Fig. 4 shows the region where initiation of fracture has taken place at the root of the cleaved flake and the crack has then cut across a series of pearlite colonies. The alternative form is when the crack passed between the consecutive cementite–ferrite layers, through the ferrite, in these cases the well known river structure was produced (Fig. 5).

When ferrite was present in the structure the mode of fracture depended upon the amount of ferrite and the heat treatment. With small quantities of ferrite, the ferrite introduced on heat treatment fractured transgranularly, no evidence of intergranular fractûre being found. As the amount of ferrite increased above 15%, the extent of transgranular fracture decreased and the remaining ferrite fractured intergranularly, usually when associated with the pearlite (Fig. 6). Finally, with a fully ferritic matrix, all the ferrite fractured intergranularly, but the fracture surface varied very much with the heat treatment: with the subcritical annealing treatment the intergranular fracture was very smooth, whereas after high temperature annealing segregation of fine precipitates to the austenite grain boundaries had taken place and fracture then occurred along the prior austenite grain boundaries (Fig. 7).

In all the fracture surfaces examined a small amount of ductile, dimple, fracture was also observed but in all cases this represented less than 5% of the total area.

Discussion

Fig. 1 shows the effect of heat treatment of the plane strain fracture toughness values. It can be seen that this parameter decreases by approximately 40% on fully heat treating the as cast material. On introducing a small amount of ferrite the fracture toughness increases slightly but decreases as the amount present exceeds 15%. In contrast, the impact value is higher in the fully heat treated material than in the as cast material.

In all materials tested, observations obtained from specimens removed from the test rig before complete separation, revealed that graphite flakes were cleaved well ahead of the main crack front. The present work confirms the presence of cleaved graphite flakes at all stress levels within the material and that deformation at the centre of the test bar is similar to that at the surface. These internal cracks also cause a deviation from the linear elastic region during testing, so that 'pop-in' is not observed in these materials. The microcracks initiate fracture within the matrix and the propagation of the crack through the material is thus controlled only by the properties of the matrix.

The fracture of grey cast irons therefore, is controlled by the distribution of the graphite and the matrix structure.

Fractographic studies have shown that the cracks in pearlite can propagate either transversely to the cementite lamellae, producing a pearlitic type fracture surface, or within a single ferrite region, in which case the river pattern is observed. Decohesion of the interface between the graphite flakes and pearlite colonies can occur, but since the initiating crack is contained inside the flake, the graphite-matrix interface is normally left intact. In pearlite, the initiation of brittle cracks appears to be relatively easy, though propagation seems more difficult and cracks are often observed to terminate in or at the boundary of pearlite regions. When initiated, the crack can only propagate a short distance before meeting obstacles in the

form of other pearlite lamellae. The crack must then propagate, either by passing round the lamellae, involving some tearing and energy absorption, or by causing cracking of the lamellae with a similar result. Thus the as cast materials can be expected to give a high toughness value because of the pearlite acting as a crack arrester.

There is a slight difference between the properties of the two pearlitic as-cast materials arising from the difference in flake size. For the material cast into larger test bars, although the flake size is itself larger so also is the interglake spacing, since the volume fraction of graphite is unaltered. As the propagation of the crack through the material is controlled by the properties of the matrix, it follows that a higher K_{Ic} value will be obtained for a slight increase in flake size.

The effect on K_{Ic} for the remaining standard and large flake cast irons is determined by the heat treatment carried out.

For the heat treatments above the critical temperature, the ferrite introduced consisted of a thin region round the graphite flakes and can be considered as introducing a small region of fine grain sized material into the structure. A study of the fracture surfaces reveals that this ferrite fractures transgranularly. The K_{Ic} value can, therefore, be expected to increase for small additions of ferrite, but decreases as the size of the ferrite becomes greater than the pearlite interlamellar spacing, at concentrations of approximately 15% ferrite. From the plot of Fig. 1 this can be seen to be true and the curve does decrease at about 15% ferrite (hardness) value in the region of 155 V.H.N. At the same time as the amount of ferrite increases to give a decrease in K_{Ic} values, so also does the mode of fracture change. The pearlite which still remains in these materials, either in the lamellar or spherodised form, fractures in the same way as already discussed and it is now the ferrite that is the controlling mechanism. The amount of transgranular fracture gradually decreases until at 100% ferrite matrix, there is only intergranular fracture, of which there are two types: fracture along the ferrite grain boundary and fracture along the prior austenite grain boundary. The fracture surfaces of this heat treated material contain a number of relatively small zones, in which decohesion has occurred along the ferrite grain boundaries. It is thought that fractures of this type are the result of grain boundary segregation, even though direct proof of this assumption is not available.

Fracture through the prior austenitė grain boundaries arises from the nature of the heat treatment. As the heat treatment involves partial reaustenitization followed by a slow cool, segregation and precipitation of possibly FeS and FesP occurs at the austenite grain boundaries. This segregation leads to an effect very similar to temper brittleness. The amount of this type of fracture increases as the amount of ferrite present increases and consequently the plane strain fracture toughness value decreases. In the fully heat treated material this low fracture toughness value is caused purely by the heat treatment, but unfortunately, when using this high temperature anneal to achieve a fully ferritic structure, a slow cooling rate is essential, to avoid pearlite formation.

The alternative heat treatment involved a subcritical anneal at 700 °C followed by an air cool. In this way the problem of retained pearlite is removed but longer times at the annealing temperatures are required. From Fig. 1, it can be seen that this form of heat treatment has a similar effect on the fracture toughness values as the alternative form. The K_{Ic} falls rapidly after a certain amount of ferrite (about 15%) has been formed. With this treatment, however, intergranular fracture through the prior austenite grain boundaries does not occur as segregation no longer takes place. Fracture at the high percentages of ferrite is now controlled by intergranular fracture through the ferrite grain boundaries.

In order to obtain reliable fracture toughness data in most materials it has been found necessary to introduce a fatigue crack into the material prior to testing in order to provide reliable crack initiation. This has been found unnecessary in the case of flake cast irons, which are inherently notch insensitive, because of the graphite flakes providing an atomically sharp crack within the material. Subsequent fatigue damage will not improve on this crack initiator.

Conclusions

1. The plane strain fracture toughness value of a standard flake cast iron (as cast) has been determined at 19·45 hectobars-cm$^{1/2}$. On heat treating to produce an annealed iron the toughness at first increases slightly but then decreases to a value only about 60% of the as cast value. The maximum K_{Ic} value corresponds to a flake cast iron with about 15% ferrite associated with the graphite flakes. The variation in K_{Ic} values depends very much on the nature of the fracture through the matrix which in turn depends on the heat treatment. The variation of fracture toughness with heat treatment of the large flake cast iron is similar to that of the standard flake iron.

2. When using a high temperature anneal followed by a furnace cool to remove the pearlite, an effect similar to temper brittleness is obtained. In the case of flake cast irons segregation of FeS or Fe_3P to the austenite grain boundaries occurs and subsequent fracture is by intergranular fracture along the prior austenite grain boundaries. This effect considerably reduces the fracture toughness of the material. When using the subcritical anneal, involving air cooling, segregation does not then occur, although the fracture toughness is still reduced.

3. In all grey cast irons fracture is shown to be initiated ahead of the main crack front by cleavage of the graphite flakes. These microcracks are linked by general fracture within either the pearlitic, ferritic or pearlitic/ferritic matrix.

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Table 1

Alloy type	Hardness V.H.N.	0·2% Yield hectobars	Tensile strength hectobars	Charpy 20°C joules	K_{Ic} hectobars-cm$^{1/2}$	% Matrix ferrite present*
Standard flake as cast	212	24·70	26·30	6·8	19·45	0
Standard flake heat treated A†	158	17·68	19·30	7·7	20·37	13·2
Standard flake heat treated B	134	15·84	17·03	8·1	15·16	59·6
Standard flake heat treated C	129	14·43	15·52	8·5	13·84	68·1
Standard flake heat treated D	124	14·41	15·50	7·1	13·26	78·0
Standard flake heat treated E	109	14·35	15·44	8·1	12·40	100·0
Large flake as cast	197	20·20	21·60	7·5	20·20	0
Large flake heat treated F	179	18·50	19·88	6·2	19·03	~1·0
Large flake heat treated G	170	18·01	19·34	6·8	19·34	4·1
Large flake heat treated H	149	17·10	18·29	7·1	18·29	26·4
Large flake heat treated I	140	16·43	17·68	7·4	16·17	40·7

*Percentages measured by line intercept method.
† Refer to experimental procedure.

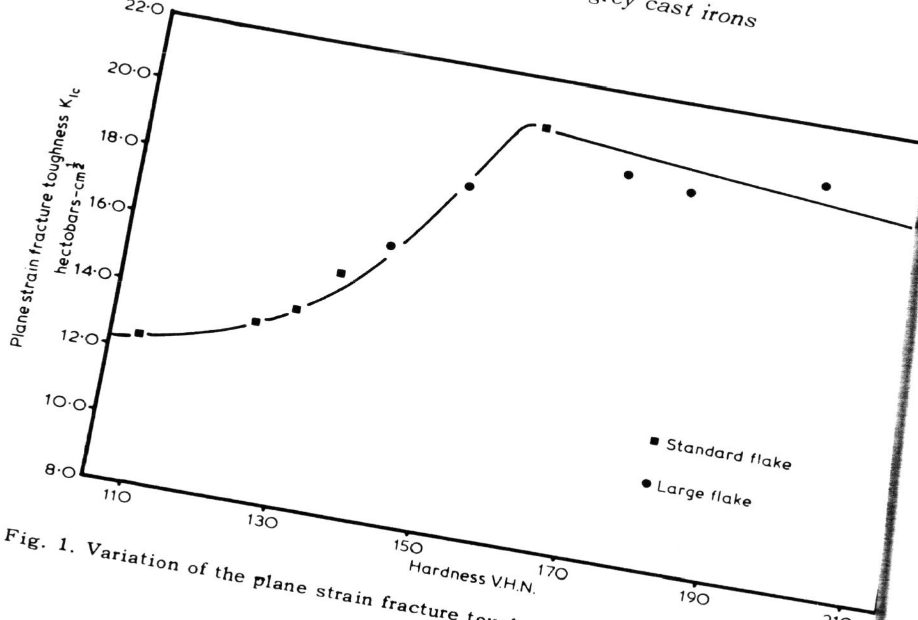

Fig. 1. Variation of the plane strain fracture toughness value (K_{Ic}) with hardness.

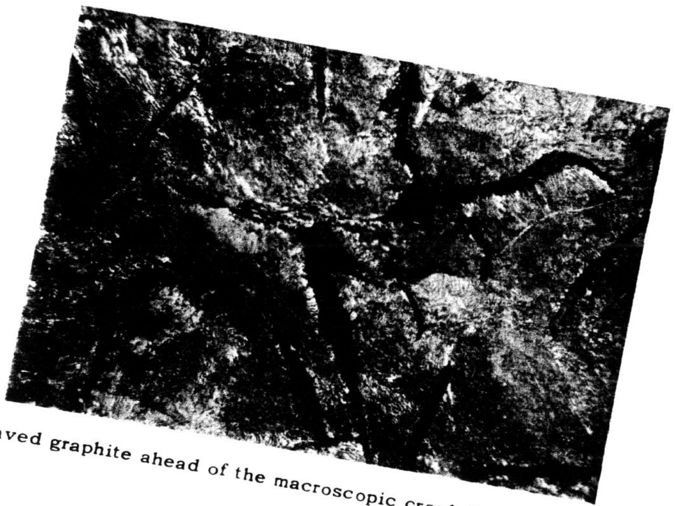

Fig. 2. Cleaved graphite ahead of the macroscopic crack front, optical × 600.

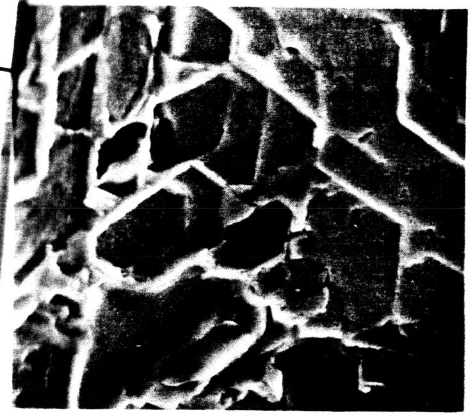

Fig. 3. Direct surface of the cleaved graphite flakes, Stereoscan × 4,100.

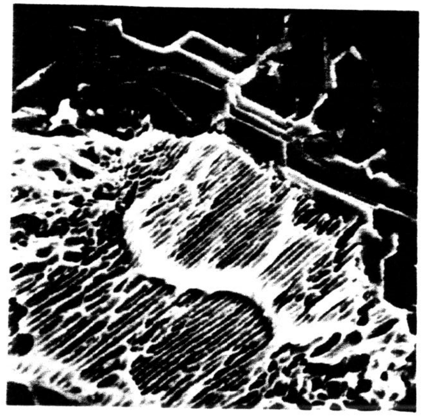

Fig. 4. Fracture of pearlite, showing etched structure, Stereoscan × 4,000.

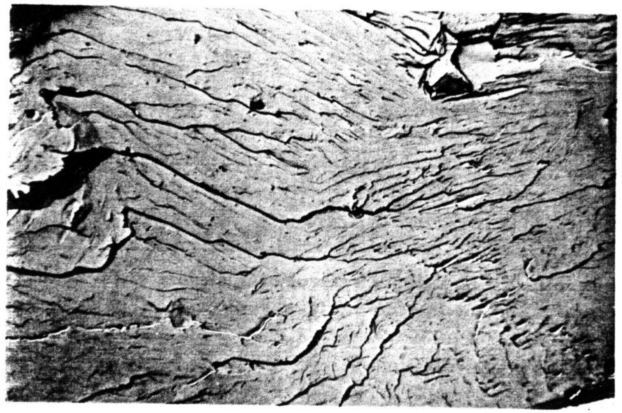

Fig. 5. River structure through ferrite in pearlite, replica × 4,000.

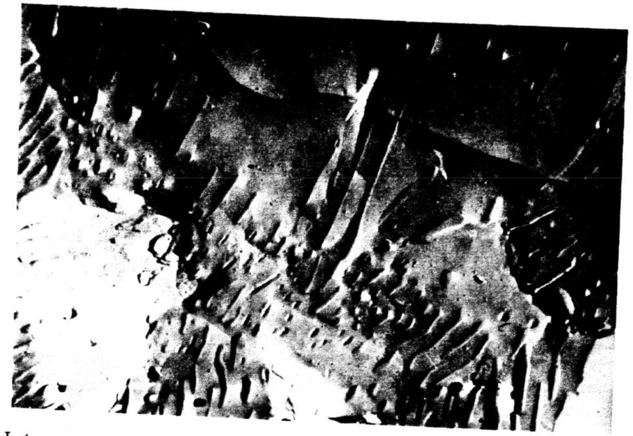

Fig. 6. Intergranular fracture through ferrite surrounded by pearlite, replica × 6,000.

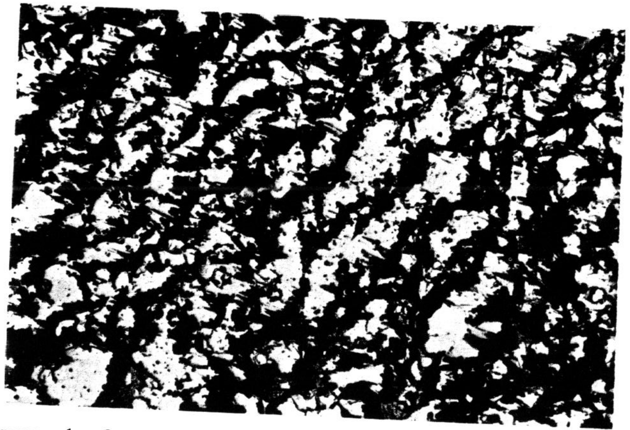

Fig. 7. Intergranular fracture through the prior austenite grain boundaries, note the fine precipitates on the fracture surface, replica × 8,000.